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# CREEP RUPTURE OF DUCTILE MATERIALS SUBJECTED TO STRAIN HARDENING OR TIME HARDENING CREEP

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Department of Aeronautics and Astronautics
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# CREEP RUPTURE OF DUCTILE MATERIALS SUBJECTED TO
# STRAIN HARDENING OR TIME HARDENING CREEP

by

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ABSTRACT

An analysis of creep rupture of ductile materials subjected to
strain hardening creep is presented. The analysis is similar to an
earlier one by Hoff, but is based on a different generalization of
the primary creep rate relation. Both approaches are then extended
to time hardening creep. Simple approximate formulae for critical
times are presented and the results of both approaches for the two
types of creep are compared. Correlation with experimental data for
some aluminum-copper, aluminum-magnesium and aluminum-zinc alloys
shows fairly good agreement of rupture times for these very ductile
materials. However, attempted correlation with data for other
materials which do not exhibit such prominent ductility indicates
that the basic assumption of the analysis, that creep rupture is
caused primarily by a process of reduction of area, holds only for
very ductile materials.

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# NOTATION

| | |
|---|---|
| $A_o$ | initial cross sectional area of specimen |
| $b$ | $(1-p)/p$ |
| $m$ | constant in primary creep law, Eq.(1), called sometimes "primary creep index" |
| $p$ | constant in primary creep law, Eq.(1) |
| $R$ | $1+\epsilon$ , defined by Eq.(9) |
| $q$ | $p-1$ |
| $s$ | $mp$ |
| $t$ | time |
| $t_{cr}$ | rupture time |
| $t_{\epsilon_{nat}}$ | rupture time predicted for strain hardening creep with Hoff's generalization |
| $t_{\epsilon_{ENG}}$ | rupture time predicted for strain hardening creep with the alternative generalization proposed here |
| $t_{TH.\epsilon_{nat}}$ | rupture time predicted for time hardening creep with Hoff's generalization |
| $t_{TH.\epsilon_{ENG}}$ | rupture time predicted for time hardening creep with the alternative generalization |
| $t_{exp}$ | rupture time from tests |
| $\beta$ | $t_{\epsilon_{nat}}/t_{\epsilon_{ENG}}$ |
| $\gamma$ | $t_{TH.\epsilon_{nat}}/t_{TH.\epsilon_{ENG}}$ |
| $\delta$ | $t_{\epsilon_{nat}}/t_{TH.\epsilon_{nat}}$ |
| $\epsilon$ | engineering strain |
| $\epsilon_{nat}$ | natural strain $= \ln(1+\epsilon)$ |
| $\eta$ | $t_{\epsilon_{ENG}}/t_{TH.\epsilon_{ENG}}$ |
| $\lambda$ | constant in primary creep law, Eq.(1) |

## NOTATION (Cont'd.)

$\sigma$    true stress

$\sigma_0$    initial applied stress

$\sigma_u$    ultimate stress of the material at test temperature

# INTRODUCTION

The creep rupture of a tensile bar of ductile material, undergoing creep deformations which are governed almost entirely by a steady state creep law, was analyzed by Hoff (Ref.1). A very simple approximate formula was obtained for the critical time, which agreed well with experimental data for pure aluminum and some aluminum alloys. However, many creep curves, especially at higher stresses and temperatures, do not exhibit the usual steady phase, but have a longer primary phase which transforms directly into a tertiary phase (see for example, Fig.1, replotted from Ref.2).

The primary phase is usually considered to be governed by a creep law of the form

$$\epsilon = (\sigma/\lambda)^m \, t^{1/p} \tag{1}$$

where $\epsilon$ is the engineering strain, $\sigma$ the applied stress, $t$ the time and $m, \lambda$ and $p$ are constants. For variable stress, Eq.(1) has to be considered as defining implicitly the strain rate as a function of the stress and the strain, or as a function of the stress and the time (Ref.3). Then, for strain hardening creep behavior, one obtains

$$\dot{\epsilon} = (1/p)(\sigma/\lambda)^{mp} \, \epsilon^{1-p} \tag{2}$$

which is essentially the relation proposed much earlier by Nadai (Ref.4).

For time hardening behavior, which is sometimes observed when time of exposure has a more pronounced effect than the amount of creep strain, a parallel relation for the primary creep rate results

$$\dot{\epsilon} = (1/p)(\sigma/\lambda)^m \, t^{(1-p)/p} \tag{3}$$

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## RUPTURE TIME FOR STRAIN HARDENING CREEP

For strain hardening creep, Hoff (Ref.5) obtained the creep rupture time by assuming that Eq.(2) can be generalized to hold for large deformations

$$\dot{\epsilon}_{nat} = (1/p)(\sigma/\lambda)^{mp} \, \epsilon_{nat}^{1-p} \tag{4}$$

where $\sigma$ is the true stress and $\epsilon_{nat}$ is the natural strain defined by

$$\epsilon_{nat} = \ell n(1+\epsilon) \tag{5}$$

and

$$\sigma = \sigma_o(1+\epsilon) \tag{6}$$

$\sigma_o$ being the initial applied stress

$$\sigma_o = P/A_o \tag{7}$$

Hoff obtained a very simple approximate formula also for strain hardening creep, which becomes, when $p$ is an integer,

$$t_{cr} \simeq [p/(pm)^p](\lambda/\sigma_o)^{pm} \tag{8}$$

Hoff's generalization of the primary creep law, Eq.(4), means that the same relation between strain rate, stress and strain is assumed to hold for the entire strain range from primary creep to rupture, provided the relation is expressed in natural strain and strain rate and true stress. In the small strain region, the natural strain practically equals the engineering strains and Eq.(4) reduces to Eq.(2). Hence the assumption implies identical constants in both equations.

However, since primary creep curves are usually given only for small strains, another approach is possible. Instead of the generalization proposed by Hoff, the observed relation between engineering strain rate and strain and stress, Eq.(2), is assumed to hold only for engineering strain rate and strain, but to be valid also in the region

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of large strains. The natural strain rate involved in the calculation
of rupture time has then to be related appropriately to the engineering
strain rate which appears in Eq. (2). It is felt that such an approach
gives one more confidence in the use of the primary creep constants
which are obtained from experiments at small strains.

In other words, the two approaches are two different extrapolations
of Eq. (2) to the region of large strains. The actual applied stress
varies, due to the reduction in area, in the same manner in both
approaches and is represented by Eq. (6). The extrapolations differ,
however, with respect to strain and strain rate.

The analysis is an extension of Hoff's work. As in Refs. 1 and 5,
a constant load tensile test is considered, and the material is assumed
to be incompressible. The symbol $R$ is also introduced again,

$$R = 1 + \epsilon \tag{9}$$

Hence

$$\epsilon_{nat} = \ln R \tag{10}$$

and

$$\sigma = \sigma_o R \tag{11}$$

For strain hardening creep the strain rate is governed by Eq. (2),
$\epsilon$ being the creep strain (the total strain less the initial elastic
and plastic strain). Rupture occurs at large strains, and hence the
natural strain rate is required for calculation of the time to rupture.
Now, Eq. (5) may be rewritten, on account of Eq. (9),

$$e^{\epsilon_{nat}} = 1 + \epsilon = R \tag{12}$$

or

$$\epsilon = e^{\epsilon_{nat}} - 1 = R - 1 \tag{13}$$

By differentiating Eq.(5) with respect to time, one obtains

$$\dot{\epsilon}_{nat} = \left[1/(1+\epsilon)\right]\dot{\epsilon}$$  (14)

Substitution of $\dot{\epsilon}$ from Eq.(2), $\epsilon$ from Eq.(13) and $\sigma$ from Eq.(11) into Eq.(14) yields

$$d\epsilon_{nat}\ dt = (1/p)(\sigma_o R/\lambda)^{m}\left[(R-1)^{1-p}/R\right]$$  (15)

For convenience let

$$q = p - 1$$  (16)

and $\quad s = mp$

Separation of variables of Eq.(15) and integration yield the critical time

$$t_{cr} = p(\lambda/\sigma_o)^s \int_1^{(\sigma_u/\sigma_o)} \left[(R-1)^q/R^s\right]dR$$  (17)

where $\sigma_u$ is the ultimate stress of the material at the test temperature.

For most materials, $p$ may be approximated by 2 or 3. For these values the critical time is given by

$$t_{cr} = 2(\lambda/\sigma_o)^s\left\{-\left[1/(2-s)\right]\left[1 - (\sigma_o/\sigma_u)^{s-2}\right] + \left[1/(1-s)\right]\left[1 - (\sigma_o/\sigma_u)^{s-1}\right]\right\}$$  (18)

for $q = 1\ (p = 2)$

or

$$t_{cr} = 3(\lambda/\sigma_o)^s\left\{-\left[1/(3-s)\right]\left[1 - (\sigma_o/\sigma_u)^{s-3}\right] + \left[2/(2-s)\right]\left[1 - (\sigma_o/\sigma_u)^{s-2}\right]\right.$$
$$\left. -\left[1/(1-s)\right]\left[1 - (\sigma_o/\sigma_u)^{s-1}\right]\right\}$$  (19)

for $q = 2\ (p = 3)$ .

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For other values of q Eq.(17) can be integrated in a similar manner to yield similar expressions for $t_{cr}$. Now, since s is generally fairly large, $(c_0/\sigma_u)^{c-3}$ and the corresponding higher powers of $(\sigma_0/\sigma_u)$ in Eqs.(18) and (19) may be neglected in comparison to unity, even when $\sigma_0$ is not far from $\sigma_u$. Eqs.(18) and (19) may therefore be simplified to

$$t_{cr} \simeq (\lambda/\sigma_0)^s \left[ 2/(s-2)(s-1) \right] \qquad \text{for} \quad p = 2 \qquad (20)$$

and

$$t_{cr} \simeq (\lambda/\sigma_0)^s \left[ 6/(s-3)(s-2)(s-1) \right] \qquad \text{for} \quad p = 3 \qquad (21)$$

The analysis can readily be extended to time hardening creep defined by the creep law of Eq.(3), by both approaches. First, generalization of Eq.(3) in the manner proposed by Hoff for strain hardening creep yields for the natural strain rate

$$\dot{\epsilon}_{nat} = (1/p)(\sigma/\lambda)^m t^{(1-p)/p} \tag{22}$$

With the quantities defined by Eqs.(9), (10), (11) and (16) and denoting

$$(1 - p)/p = b \tag{23}$$

one can rewrite Eq.(22) as

$$d\ell nR/dt = (1/p)(\sigma_o R/\lambda)^m t^b \tag{24}$$

Separation of variables and integration yields

$$\begin{aligned} t_{cr} &= \left\{\left[(b+1)p/m\right](\lambda/\sigma_o)^m\left[1 - (\sigma_o/\sigma_u)^m\right]\right\}^{1/(b+1)} \\ &= (1/m)^p(\lambda/\sigma_o)^s\left[1 - (\sigma_o/\sigma_u)^m\right]^p \end{aligned} \tag{25}$$

since

$$b + 1 = 1/p \tag{26}$$

Now, if $\sigma_u \gg \sigma_o$

$$\left[1 - (\sigma_o/\sigma_u)^m\right]^p \approx 1 - p(\sigma_o/\sigma_u)^m \approx 1 \tag{27}$$

provided m is sufficiently large. As p can be taken 2 or 3 for most materials and m is usually larger than 3, the approximation of Eq.(27) does, in general, not involve large errors. With this approximation Eq.(25) becomes

$$t_{cr} \approx (1/m^p)(\lambda/\sigma_o)^s \tag{28}$$

However, it should be noted that the approximation here is inferior to that employed in the case of strain hardening creep. There, at most $(\sigma_o/\sigma_u)^{pm-3}$ was neglected in comparison with unity, whereas in Eq.(27) $p(\sigma_o/\sigma_u)^m$ is neglected in comparison with unity. Hence the error implied by Eq.(28) is $p(\sigma_u/\sigma_o)^{mp-m-3}$ times as large as that introduced by the approximation leading to Eqs.(8) or (20) and (21).

In the second approach, Eq.(3) is again assumed to hold for the engineering strain rate for both small and large strains. With the quantities defined by Eqs.(9), (10), (11), (16) and (23) and with Eqs.(3) and (14), the natural strain rate can be rewritten as

$$d\epsilon_{nat}/dt = (1/pR)(\sigma_o R/\lambda)^m t^b \tag{29}$$

Separation of variables and integration yields then, on account of Eq.(26),

$$t_{cr} = \left[1/(m-1)\right]^p (\lambda/\sigma_o)^s \left[1 - (\sigma_o/\sigma_u)^{m-1}\right]^p \tag{30}$$

For $\sigma_u \gg \sigma_o$ , an approximate expression can again be obtained for the critical time provided $m$ is sufficiently large,

$$t_{cr} \simeq \left[1/(m-1)^p\right](\lambda/\sigma_o)^s \tag{31}$$

The accuracy of this approximation is, however, even lower than that of Eq.(28), since here $p(\sigma_o/\sigma_u)^{m-1}$ is neglected in comparison with unity. The error introduced by Eq.(31) relative to Eq.(30) is therefore $(\sigma_u/\sigma_o)$ times that introduced in Eq.(28).

It is of interest to compare the rupture times obtained by the various approaches. If one compares the approximate formulae for the critical times, very simple relations are obtained. However, as the approximations in these formulae are different, the more exact formulae for the rupture time have to be employed in the comparison for the low values of m , when the initial applied stress approaches the ultimate stress and the approximations lose their validity. Only the simple approximate relations are given below, but the results of computations with the more exact relations are also shown in Figs. 2 and 3.

If β denotes the relation of the rupture time for strain hardening obtained by Hoff to that obtained by the alternative generalization proposed in this paper,

$$\beta = (s-1)(s-2)/s^2 \qquad \text{for} \quad p = 2 \qquad (32)$$

and

$$\beta = (s-1)(s-2)(s-3)/s^3 \qquad \text{for} \quad p = 3 \qquad (33)$$

if the approximate formulae, Eqs.(8) and (20), (21) are compared. Fig. 2 shows a plot of β versus the primary creep index m . For two typical initial applied stress to ultimate stress ratios, $(\sigma_0/\sigma_u)$ = 0.5 and $(\sigma_0/\sigma_u)$ = 0.9 , the values of β as obtained from a comparison of the more exact Eq.[59] of Ref.5 and Eqs.(18) and (19) are also shown.

If one compares the results of the two approaches for time hardening creep, the approximate formulae Eqs.(28) and (31) are related by

$$\gamma = \left[(m-1)/m\right]^p \qquad (34)$$

This ratio is plotted in Fig. 3 for p = 2 and p = 3 . The relation of the more exact formulae Eqs.(25) and (30) is also included for $(\sigma_0/\sigma_u)$ = 0.5 and $(\sigma_0/\sigma_u)$ = 0.9 .

Figures 2 and 3 show that as the primary creep index $m$ increases, the difference between the two approaches diminishes both for strain hardening and time hardening creep. For low applied stresses, say, $\sigma_o \leq 0.5 \, \sigma_u$ , the noticeable difference btween the two approaches, predicted by the approximate relations, is verified by the results obtained by the more exact relations. When the applied stress approaches the ultimate stress, for example, when $\iota_o = 0.9 \, \sigma_u$ , the difference practically disappears. This would be expected since the strains at rupture tend to be small when the applied stress approaches the ultimate.

Similar simple relations can be obtained by comparison of the approximate rupture time for strain hardening creep with that for time hardening creep by one of the approaches. For the first approach, which assumes that the primary creep relations hold for the entire strain range if expressed in natural strain rates and strains, the rupture time for strain haruening creep divided by that for time hardening creep yields

$$\delta = p!/p^p \tag{35}$$

For the second approach, which assumes the primary creep relations to hold only for engineering strain rates and strains, but to be valid also in regions of large strains, the rupture time for strain hardening creep divided by that for time hardening creep yields

$$\eta = (m-1)/(2m-1) \qquad \text{for } p = 2 \tag{36}$$

and

$$\eta = 2(m-1)^2/(3m-2)(3m-1) \qquad \text{for } p = 3 \tag{37}$$

However, these relations, Eqs.(35), (36) and (37), are again valid only within the accuracy of the approximations, on account of the considerable difference in the error included in the approximate formulae.

The comparison shows that the second approach yields longer critical times than the first one, in both types of creep deformation. Also time hardening creep will yield longer rupture times than strain

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hardening creep by both approaches. However, since the analysis is basically a generalization of empirical primary creep laws, only correlation with experimental data can determine the relative usefulness of the proposed formulae. This is attempted in the next section.

## CORRELATION WITH EXPERIMENTAL DATA AND DISCUSSION

The results of tests carried out by Robinson, Tietz and Dorn (Ref.2) were compared with the theoretical predictions by the two approaches for strain hardening and time hardening creep. The test materials were a series of high purity binary alpha solid solutions in aluminum. Three groups of aluminum alloys were compared: copper-aluminum alloys, magnesium-aluminum alloys and zinc-aluminum alloys. Each group includes four alloys with different percentages of alloying element, and all tests were at $300^{\circ}F$. In Tables 1, 2 and 3 the predicted and experimental values of the rupture times are given for different applied stresses. The data for the magnesium-aluminum alloys are also plotted in Fig.4.

To obtain a theoretical rupture time, the primary creep constants p, m and $\lambda$ have first to be found from the creep curves of Ref.2. The left hand portions of the creep curves (for relatively short times) have to be used to find these constants, since for longer times the curves include the effects of the reduction in area. The time index p is found from Eq.(1) with $\sigma$ and $\lambda$ constant. For the alloys of Figs. 4-6 of Ref.2, the nearest integer for p is 2. Now, from short time stress-strain curves at $300^{\circ}F$, Figs.15-17 of Ref.2, the initial strains are found for the nominal applied stresses. The total strain less the initial strain is the creep strain. The creep strains at particular times are plotted versus the corrected applied stress on a log-log scale yielding straight lines whose slope is m. The integer nearest to the average value of m obtained from about 5 or 6 of such isochronous lines is taken as the appropriate value. $\lambda$ is then found from Eq.(1) with these p and m from about 10 points and averaged.

In Tables 1, 2 and 3, the rupture time for strain hardening creep with Hoff's generalization is denoted $t_{\epsilon_{nat}}$, and that obtained with the alternative generalization of this paper $t_{\epsilon_{ENG}}$. The rupture times for time hardening creep are similarly denoted $t_{TH.\epsilon_{nat}}$ and $t_{TH.\epsilon_{ENG}}$,

respectively. The more exact formulae, Eqs. (18), (25) and (30), and an expression which is equivalent to Eq. (59) of Ref. 5, have been used in the calculation of the rupture times.

Fairly good agreement is found for most of the tests. The rupture times for time-hardening creep, with the generalization of the primary creep law proposed in this paper, appear to be closest to the experimental results. Better agreement is found for the magnesium-aluminium and copper-aluminum alloys (except the 0.101% Cu alloy for which the predictions are noticeably too long) than for the zinc-aluminum alloys. It is interesting to note that the rupture times computed with the assumption of strain-hardening creep are rather conservative for all the test results (except those for two tests of 0.101% Cu Al alloy). It might be added, that if the plastic strain component is taken into account (Ref. 6) the predicted rupture time is further shortened. Since this effect would be most noticeable at the high applied stresses, where the predictions based on time-hardening creep often exceed the experimental rupture times, it would tend to improve the overall agreement of the time-hardening theory. It appears, therefore, that for the materials compared here, a time-hardening creep law provides a realistic description of the rupture behavior.

An extensive literature search for additional experimental data of primary creep and rupture for ductile materials revealed a remarkable scarcity of such data. Most investigators report either details of primary creep behavior or rupture times, but very seldom both. Since only creep curves which do not exhibit a significant steady phase are suitable for comparison with the present theory, no additional data was found. However, through the courtesy of the Research Division of High Duty Alloys Ltd., Slough, England, creep curves for RR 58 and RR 257 aluminium alloys (which are used in the Anglo-French supersonic transport) were obtained with records of rupture times and corresponding total strains (Ref. 7). Some of these curves had no significant secondary phase. Correlation was attempted with some of these curves, but the predicted rupture times greatly exceeded the experimental values (the computed values were 50 to 30,000 times the experimental ones). As a

check on the order of magnitude of the prediction, the rupture times
were computed by the steady creep analysis of Ref.1 (the steady creep
constants were obtained from approximate tangents to the central portion
of the creep curves). However, again the predictions were many times
too large, though to a lesser extent (6 to 60 times).

During the literature search some creep curves of suitable form
were encountered, for which rupture times were also reported. However,
they represented materials which exhibited only small amounts of perma-
nent deformation prior to fracture, and hence cannot be adequately des-
cribed by the present theory. For comparison, correlation was attempted
for a typical case of this group, a 5% molybdenum steel at $1020^{\circ}F$
(Ref.8). The predicted rupture times greatly exceeded the experimental
values (by 400 to 500,000 times), and a steady creep analysis produced
similar discrepancies (700 to 3000 times).

It may be pointed out that "ductile theory", which assumes rupture
by reduction of area, disagrees noticeably with experiments also for
creep with a predominant secondary phase in the case of age-hardening
aluminium alloys. For example, for 7075 TC at $375^{\circ}F$ (Ref.5) rupture
times 10 to 20 times the experimental ones are predicted.

Hence it is apparent that theories which are based on the premise
that rupture is caused primarily by a necking process are applicable
to very ductile materials only (like those tested in Ref.2 or the 3003
and 5052 aluminium alloys tested in Ref.9). For other materials one has
to turn to other theories which try to account for brittle or partly
brittle behavior at rupture (see, for example, Refs.10, 11 or 12).

It may be observed, that the difference between the rupture times
obtained from four different creep laws, is insignificant compared to
the difference between predicted and experimental values, except for the
very ductile materials of Ref.2. Hence, the marked disagreement between
theory and experiment for the more commonly used alloys cannot be
attributed to inadequacies in the creep laws, but rather to the character
of the rupture process.

## ACKNOWLEDGMENT

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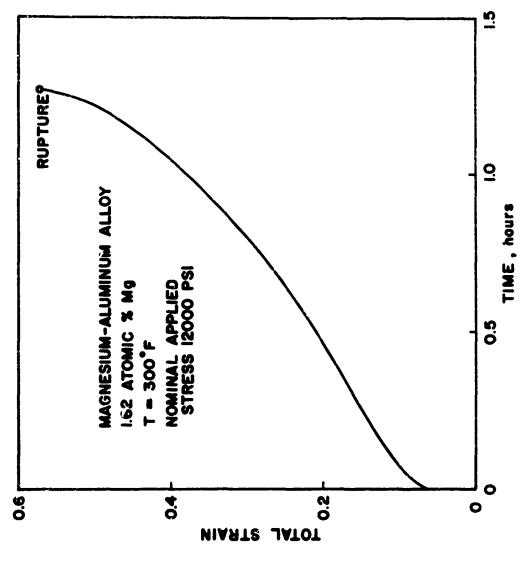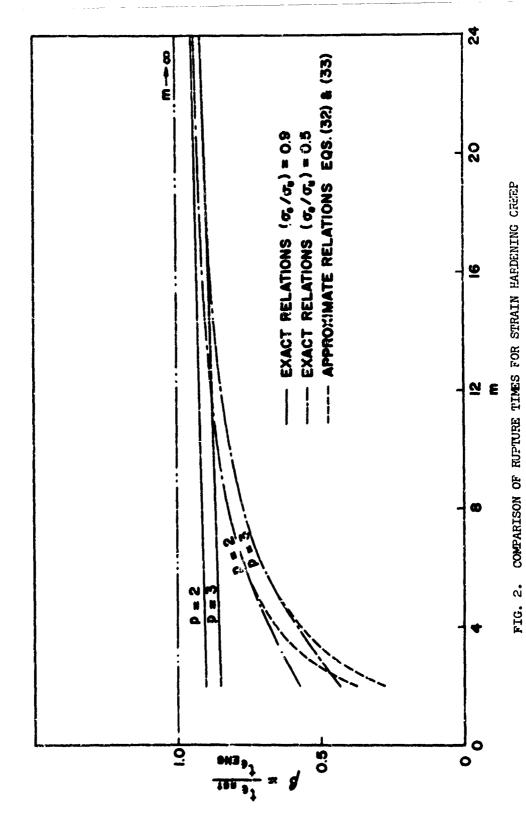
FIG. 1. TYPICAL CREEP CURVE WITHOUT SECONDARY CREEP
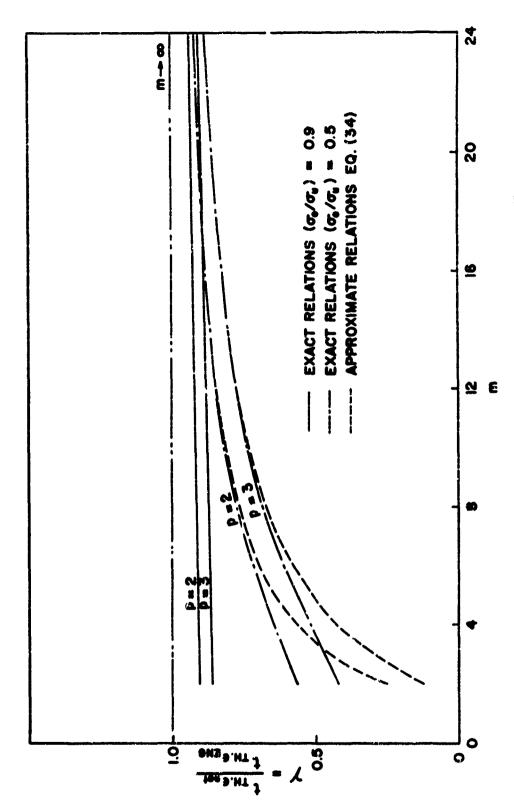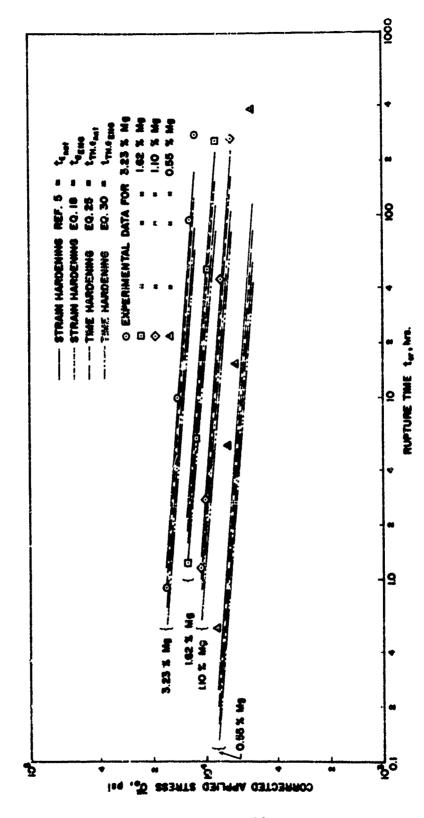(replotted from Ref.2)

MAGNESIUM-ALUMINUM ALLOY
1.62 ATOMIC % Mg
T = 300°F
NOMINAL APPLIED
STRESS 12000 PSI

RUPTURE

FIG. 2. COMPARISON OF RUPTURE TIMES FOR STRAIN HARDENING CREEP

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FIG. 3. COMPARISON OF RUPTURE TIMES FOR TIME HARDENING CREEP

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FIG. 4.- COMPARISON OF THEORY WITH TESTS ON ALUMINUM
MAGNESIUM ALLOYS BY ROBINSON, TIETZ and DORN

(Ref.2) .

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## TABLE 1

### Comparison of Theory and Experiment: Copper-Aluminum Alloys at 300°F Tested by Robinson, Tietz and Dorn

| Atomic % of Alloying Element | Nominal Appl. Stress (psi) | Predicted Rupture Times (Hours) | | | | Experimental Rupture Times $t_{exp.}$ (Ref.2) |
| | | Strain Hardening (Ref.5) | | Time Hardening | | |
| | | $t_{\epsilon_{nat}}$ | $t_{\epsilon_{ENG}}$ | $t_{TH.\epsilon_{nat}}$ | $t_{TH.\epsilon_{ENG}}$ | |
|---|---|---|---|---|---|---|
| 0.029% Cu | 2800 | 29.7 | 50.9 | 54.7 | 102 | 154 |
| | 3500 | 6.71 | 10.9 | 11.5 | 19.7 | 26.4 |
| | 4000 | 2.56 | 4.12 | 4.12 | 6.56 | 6.73 |
| | 4800 | 0.48 | 0.66 | 0.67 | 0.93 | 1.00 |
| 0.054% Cu | 2900 | 61.1 | 92.7 | 119 | 201 | 216 |
| | 3500 | 12.6 | 18.9 | 23.9 | 38.9 | 49.4 |
| | 4000 | 4.01 | 5.94 | 7.37 | 11.6 | 13.0 |
| | 5000 | 0.51 | 0.71 | 0.83 | 1.19 | 2.12 |
| 0.101% Cu | 3000 | 501 | 696 | 995 | 1533 | 380 |
| | 3500 | 104 | 144 | 205 | 312 | 100 |
| | 4000 | 25.8 | 36.5 | 51.1 | 76.6 | 38.0 |
| | 6000 | 0.27 | 0.35 | 0.44 | 0.59 | 1.01 |
| 0.233% Cu | 4000 | 634 | 881 | 1257 | 1929 | >1000 |
| | 4800 | 96.0 | 156 | 191 | 289 | 220 |
| | 5500 | 23.7 | 32.6 | 45.3 | 67.0 | 41.0 |
| | 6500 | 3.86 | 4.06 | 7.03 | 9.95 | 7.60 |
| | 7500 | 0.70 | 0.93 | 1.18 | 1.57 | 1.28 |

## TABLE 2

### Comparison of Theory and Experiment: Magnesium-Aluminum Alloys at 300°F Tested by Robinson, Tietz and Dorn

| Atomic % of Alloying Element | Nominal Appl.Stress (psi) | Predicted Rupture Times (Hours) | | | | Experimental Rupture Times $t_{exp.}$ (Ref.2) |
| --- | --- | --- | --- | --- | --- | --- |
| | | Strain Hardening | | Time Hardening | | |
| | | $t_{\epsilon_{rat}}$ (Ref.5) | $t_{\epsilon_{ENG}}$ | $t_{TH.\epsilon_{rat}}$ | $t_{TH.\epsilon_{ENG}}$ | |
| 0.55% Mg | 5300 | 42.4 | 55.5 | 32.6 | 116 | 385 |
| | 6500 | 3.17 | 4.08 | 5.84 | 7.87 | 15.5 |
| | 7200 | 0.73 | 0.93 | 1.27 | 1.64 | 5.44 |
| | 8000 | 0.13 | 0.16 | 0.20 | 0.24 | 0.54 |
| 1.10% Mg | 7000 | 82.0 | 107 | 163 | 234 | 266 |
| | 8000 | 14.6 | 19.1 | 28.7 | 40.7 | 45.0 |
| | 9500 | 1.40 | 1.82 | 2.50 | 3.44 | 2.78 |
| | 10000 | 0.60 | 0.78 | 0.99 | 1.33 | 1.16 |
| 1.62% Mg | 8500 | 82.0 | 112 | 154 | 224 | 260 |
| | 9400 | 28.1 | 38.2 | 50.8 | 71.8 | 50.8 |
| | 10800 | 5.87 | 7.65 | 9.71 | 13.0 | 6.00 |
| | 12000 | 1.31 | 1.61 | 1.92 | 2.38 | 1.23 |
| 3.23% Mg | 11300 | 65.0 | 85.0 | 126 | 177 | 280 |
| | 12000 | 29.7 | 38.7 | 57.2 | 79.1 | 95.4 |
| | 14000 | 3.82 | 4.93 | 7.04 | 9.38 | 10.1 |
| | 16000 | 0.57 | 0.70 | 0.92 | 1.16 | 0.91 |

## TABLE 3

Comparison of Theory and Experiment: Zinc-Aluminum
Alloys at 300°F Tested by Robinson, Tietz and Dorn

| Atomic % of Alloying Element | Nominal Appl.Stress (psi) | Predicted Rupture Times (Hours) | | | | Experimental Rupture Times (Ref.2) $t_{exp.}$ |
| | | Strain Hardening (Ref.5) | | Time Hardening | | |
| | | $t_{\epsilon_{nat}}$ | $t_{\epsilon_{ENG}}$ | $t_{TH.\epsilon_{nat}}$ | $t_{TH.\epsilon_{ENG}}$ | |
|---|---|---|---|---|---|---|
| 0.21% Zn | 2500 | 28.3 | 42.3 | 54.3 | 89.7 | 268 |
| | 3000 | 5.86 | 8.69 | 10.8 | 17.1 | 36.3 |
| | 3400 | 1.87 | 2.70 | 3.28 | 4.97 | 8.00 |
| | 4000 | 0.33 | 0.44 | 0.51 | 0.71 | 0.31 |
| 0.40% Zn | 2500 | 22.7 | 35.7 | 43.6 | 72.0 | 200 |
| | 3000 | 4.62 | 6.63 | 8.68 | 13.7 | 25.6 |
| | 3400 | 1.49 | 2.13 | 2.63 | 3.97 | 6.77 |
| | 4000 | 0.26 | 0.35 | 0.41 | 0.56 | 1.20 |
| 0.76% Zn | 2500 | 25.9 | 39.2 | 49.9 | 82.7 | 206 |
| | 2800 | 9.81 | 14.7 | 18.5 | 29.9 | 42.2 |
| | 3400 | 1.82 | 2.64 | 3.22 | 4.89 | 8.00 |
| | 4000 | 0.33 | 0.45 | 0.53 | 0.74 | 1.29 |
| 1.62% Zn | 2300 | 70.9 | 98.4 | 141 | 217 | 339 |
| | 2500 | 30.5 | 42.3 | 60.3 | 92.3 | 155 |
| | 2800 | 9.63 | 13.4 | 18.9 | 28.7 | 50.2 |
| | 3400 | 1.22 | 1.69 | 2.34 | 3.44 | 8.00 |
| | 4000 | 0.20 | 0.27 | 0.37 | 0.52 | 1.70 |

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